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20. ABSTRACT (Continue on reverse side if necessary and identify by block number)

Recovery processes in the form of clustering of dislocation configurations in both plastically deforming or creeping copper and nickel single crystals have been investigated by conventional electron microscopy and by high voltage electron microscopy by means of in-situ experiments. The creep response of such alloys and the stages of the recovery processes have also been modeled.

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MECHANISM OF RECOVERY IN PLASTIC DEFORMATION AND CREEP

FINAL REPORT

(NO. 1)

Ali S. Argon

July 23, 1982

U.S. ARMY RESEARCH OFFICE (DURHAM)

CONTRACT NO. DAAG-29-78-C-0014

MECHANICAL ENGINEERING DEPARTMENT
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I. Statement of the Problem Investigated

The investigation carried out under support from the ARO, which is reported here and in more detail in the several published papers, was performed for the purpose of developing a better understanding of the mechanisms of recovery in plastic deformation and creep. The investigation was both of an experimental and of a theoretical nature.

II. Principal Results of the Research

2.1 Introduction

The accurate solution of deformation processing problems in the intermediate temperature range requires the statement of the current deformation resistances of the solid as a tensor quantity, and the laws of evolution of these resistances as a function of the current state, stress, temperature and time -- in an incremental form. Many useful approximations to the actual behavior of a deforming solid exist and have been used in idealized form by applied mechanicians and materials scientists. Other more sophisticated forms of behavior have been proposed and have been used in special applications for modeling forming at elevated temperatures and fracture processes on creeping alloys. A central weakness of all such models is the inadequacy of the state of understanding of processes of thermal and dynamic recovery that affect the creep resistance of alloys at intermediate and at elevated temperatures. The goal of the research that is being summarized here has been the elucidation of these processes from a mechanistic point of view rather than the more prevalent point of view of describing this behavior by abstract internal parameters.

2.2 Recovery Mechanisms in Creep

In all pure metals and in many single phase alloys the dislocation density alone is not an adequate parameter to describe the creep resistance. It is necessary also to take note of the clustering of the retained dislocations into sub-grain walls. The creep resistance of an alloy is inversely related to the sub-grain size and that manipulations of this dimension to stably reduce it by any means possible produces desirable improvements in creep resistance. At steady state creep the average dimension of sub-grains remains constant. It has been stated that when the applied stress is changed to evoke a new steady state the sub-grain dimension will reversibly change to adjust itself to the new conditions. However, this reversibility has only been demonstrated partially, and then only in aluminum [1]. Nevertheless, the sub-grain size remains the most prominent state parameter in creep.

The clustering of dislocations into sub-grain walls requires their ready repositioning in space to reduce their interaction energy. This is accomplished by a combination of cross glide and climb. While the former has been relatively well understood [2, 3] for both unextended and extended dislocations, the latter has been previously treated with any precision only for unextended dislocations [4]. Since many pure metals and alloys of interest have relatively low stacking fault energy it was found essential to understand in some detail the climb process of extended dislocations.

This problem has been analysed by Argon and Moffatt [5] who demonstrated the possibility for the climb of extended jogs along extended edge dislocations without constrictions. The analysis has led naturally to a recognition that the climb process is vacancy-evaporation limited and that the climb rate becomes proportional to the inverse square of the width of the stacking fault of the extended edge dislocation, explaining directly an important aspect of stacking fault strengthening of creeping alloys.

2.3 Internal Stresses in Subgrain Forming Alloys

Many investigators have recognized that pure metals and so-called Class II alloys form sub-grains in later stages of creep approaching steady state, and that in such metals and alloys substantial dynamic internal stresses can be measured by stress-dip tests [6]. These alloys are also distinguished from Class I alloys by a stress exponent of 4.5-5.5 of the "steady state" creep rate. It has been proposed by Argon and Takeuchi [7] that this internal stress is a result of bowing under stress of migrating sub-grain boundaries, and that the strain-producing glide dislocaions respond to the effective stress obtained by subtracting the dynamic internal stress from the applied stress. A careful analysis of the steady state dislocation flux subject to the random internal stress and incorporating climb steps for these dislocations during temporary storage at the "holding sites" around the subgrain boundaries leads naturally to a dependence of the minimum creep rate on the third power of the stacking fault energy observed by many investigators, and a stress exponent of between 4.5-5.5. In addition, the theoretical model of Argon and Takeuchi [7] gives a very satisfactory accounting of the reported levels of the measured internal stresses in alloys, and is compatible with the creep transients studied by Gibeling and Nix [8].

2.4 Steady State Creep in Sub-grain Forming Alloys

The evolutionary processes that lead to sub-grain formation near steady state creep have been analysed and modeled by Argon, Prinz and Moffatt [9], Prinz, Argon and Moffatt [10], and Gottstein and Argon [11]. In these models dislocations are considered generated at sources around the sub-grain boundaries where the issuing of a dislocation from a typical source is subject to a waiting period that increases with the width of extension of edge dislocations. Dislocations released from sources interact with other glide dislocations moving in the opposite direction (inter-plane hardening) and must move through the field of internal stress undulation produced by the randomly flexed sub-boundaries under the applied stress. The glide dislocations as dipolar pairs go with temporary storage near slowly migrating subgrain boundaries where they make up the redundant population of the subgrain boundary giving these boundaries a finite dynamic thickness. These redundant dislocations gradually break-up by pinching-off events, go through a stage of loop-debris before the debris is finally eliminated by climb governed by core diffusion [9, 10]. The loop-debris observed as an ubiquitous feature of dynamic sub-grain boundaries is found to have low glide mobility and has been identified as the primary limitation to the migration mobility of subgrain boundaries [9, 10]. The basic explanation of this lack of glide mobility of the loop debris which is primarily constituted of prismatic dislocations is not clear. Thus, the typical sub-grain boundary contains: a) a set of geometrically necessary dislocations, most prominently consisting of two parallel sets that always have a unique direction for migration in a self-similar manner without the need of diffusion; b) a collection of redundant dislocation as dipolar pairs driven into the boundaries and in all stages of decomposition; and c) loop-debris as the principal product of decomposition of redundant dislocations, and also as the principal source of drag to the migration mobility of sub-boundaries. The basic rate limiting step that maintains the boundary in a self similar manner at a given state of "fluffiness" has been found to be core diffusion, by means of in-situ annealing experiments carried out in the HVEM at the Argonne National Laboratory [10].

2.5 Steady State Creep

Although the notion of a steady state where both kinetic processes are in equilibrium and where the defect micro-structure exists in a stationary state, on the average, has been widely accepted, the assumed structural reversibility with change in stress has never been demonstrated. In experiments specifically designed to demonstrate this possibility Langdon [1] has found only a reversibility in the mechanical conditions of the steady state without reversibility in the defect micro-structure.

In creep experiments on single crystals of copper and nickel carried out during the course of this investigation steady state conditions of creep could never be achieved in the temperature range below 0.8 T_m. In every such instance steady state was preceded by re-crystallization. Only in experiments on copper single crystals carried out above 0.8 T_m could steady state be achieved without re-crystallization. A survey of the literature has reinforced these observations. Except in aluminum where reliable observations on steady state creep exist, in other metals and alloys the reported deformation appears to be related to a minimum creep rate in polycrystals where the reduction in strain rate due to hardening is offset by an increase in cavitational strain rate. In view of this the majority of the published creep information on polycrystalline alloys becomes suspect and must be treated in a special light. True, non-dilatational steady state creep rates, if they exist, must be smaller in every case.

III. Participating Scientific Personnel

Ali S. Argon, Professor of Mechanical Engineering, principal investigator.

Friedrich Prinz, Visiting Associate Professor of Mechanical Engineering, researcher.

Gunter Gottstein, Visiting Associate Professor of Materials Science and Engineering, researcher.

William Moffatt, NSF Fellow, IBM Fellow, Materials Science and Engineering.

William Whitelow, graduate student, research assistant.

IV. Degrees Earned

None

V. Technical Publications

1. "Climb of Extended Edge Dislocations," by A.S. Argon and W.C. Moffatt, Acta Metallurgica, vol. 29, pp. 293-299 (1981).
2. "Internal Stresses in Power-Law Creep," by A.S. Argon and S. Takeuchi, Acta Metallurgica, vol. 29, pp. 1877-1884 (1981).

3. "Dislocation Creep in Subgrain-Forming Pure Metals and Alloys," by A.S. Argon, F. Prinz, and W.C. Moffatt, in Creep and Fracture in Engineering Materials and Structures, edited by B. Wilshire, and D.R.J. Owen, Pineridge Press: Swansea, U.K., pp. 1-15 (1981).

4. "Recovery of Dislocation Structures in Plastically Deformed Copper and Nickel Single Crystals," by F. Prinz, A.S. Argon, and W.C. Moffatt, Acta Metallurgica, vol. 30, pp. 821-830 (1982).

5. "A Kinetic Dislocation Model for the Plastic Behavior at High Strains," by F. Prinz, A.S. Argon, and W.C. Moffatt, submitted to Materials Science and Engineering.

References Quoted in the Report

1. T.G. Langdon, R.B. Vastava, and P. Yavari, in Strength of Metals and Alloys, edited by P. Haasen, V. Gerold, G. Kostorz, Pergamon Press: Oxford, vol. 1, pg. 271 (1979).

2. G. Schoek, and A. Seeger, in "Report of the Conference on Defects in Solids," Physical Society: London, p. 340 (1955).

3. M.J. Marienkowski, K. Sadananda, and M.J. Olson, Crystal Lattice Defects, 5, 187 (1974).

4. J. Friedel, Dislocations, Addison Wesley: Reading, Mass., pg. 104 (1964).

5. A.S. Argon and W.C. Moffatt, Acta Met., 29, pg. 293 (1981).

6. S. Takeuchi and A.S. Argon, J. Mater. Sci., 11, 1542 (1976).

7. A.S. Argon and S. Takeuchi, Acta Met., 29, 1877 (1981).

8. J.C. Gibeling and W.D. Nix, Acta Met., 29, 1769 (1981).

9. A.S. Argon, F. Prinz, and W.C. Moffatt, in Creep and Fracture in Engineering Materials and Structures, edited by B. Wilshire, and D.R.J. Owen, Pineridge Press: Swansea, U.K., pg. 1 (1981).

10. F. Prinz, A.S. Argon and W.C. Moffatt, Acta Met., 30, 821 (1982).

11. G. Gottstein and A.S. Argon, to be submitted to Acta Metallurgica.

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